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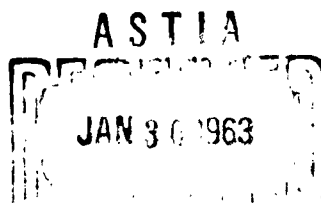
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DEFORMABLE ALUMINUM ALLOYS
(SELECTED ARTICLES)



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DEFORMABLE ALUMINUM ALLOYS (SELECTED ARTICLES)

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TABLE OF CONTENTS

| | PAGE |
|--|------|
| Features of the Thermal Treatment of Heavy Forgings of D1 Alloy, by V. I. Isayev, I. A. Ivankin, et. al. | 1 |
| Hot Cracking of Alloys of the V95 Type, by I. I. Novikov and A. E. Semenov | 8 |
| Some Features of Obtaining Ingots From the Alloys AMg5V and AMg6, by V. A. Livanov, R. M. Gabidullin and L. P. Donorskaya | 15 |

FEATURES OF THE THERMAL TREATMENT OF HEAVY
FORGINGS OF D1 ALLOY

V. I. Isayev, I. A. Ivankin, V. I. Kulakov,
and N. A. Loktionova

In mechanical treatment of heavy drop forged articles of deformable aluminum alloys one of the most widespread types of spoilage is warping or the formation of cracks in them.

Rapid cooling and non-uniform removal of heat lead to non-uniform volumetric changes in the profile of the articles, resulting in large residual thermal stresses.

The effects of these stresses come to light soon after hardening — the articles change their dimensions — they warp, or cracks appear in them. Oftenest of all, the effect of residual stresses as a result of redistribution comes to light in mechanical treatment. The nature of the redistribution depends on the shape of the article, the sequence of the operations in physical treatment, the quality of the cutting instrument, and other factors.

The most effective way to lower the residual stresses is to decrease the rate of cooling when quenching articles made of aluminum alloys in the 535-200° temperature range, since at these temperatures

occur basic volumetric changes. Lowering the cooling rate to $2-3^{\circ}$ per second in the indicated temperature range secures a small temperature gradient across the profile and the fixation of a supersaturated solid solution capable of strain hardening in subsequent aging, and also secures a sufficiently high level of physical properties and a sharp decrease in residual stresses.

In the present article based on research done and the processing of statistical data on the manufacture of forgings in batch production are described some peculiarities of the thermal treatment of heavy forgings of D1 alloy; and the effect of the individual factors in heat treatment on the magnitude of the originating residual stresses and the level of the mechanical properties is ascertained.

Forgings were made of poured homogenized billets by upsetting them by hammer along the axis of the ingot and finally drop-forging them. The temperature regimes for forging and drop-forging correspond to the technology of batch production for D1 alloy. For investigating the structure and physical properties in different sections specimens were cut out of the forging in the pattern in Figure 1. The part of the forgings from which the billets for the specimens were cut out was heated in a $500^{\pm 30}$ saltpeter bath (exposure at this temperature 2 hrs.) then it was quenched in 20° water and subjected to natural aging for 96 hours. The results of testing the physical properties of the specimens cut out of the various zones of the forging are listed in Table 1.

The results of research on several regimes of heat treatment (Table 2) showed that heating to a quenching point of $490^{\pm 30}$ with exposure for 2 hours and cooling in water at 70% secures a decrease in spoilage from cracks to 0.2%. The cracks were discovered by an

ultrasonic defectoscope; and in the macrosection, in the form of fine hairline cracks.

An analysis of the production data on the formation of cracks in forgings in respect to the regimes of heat treatment confirmed the linkage between the cracks and the temperature of heating to the quenching point and the temperature of the water in the quenching tank.

In the literature there are also indications of a similar relation of spoilage from cracks to the temperature of heating to the quenching point in articles made of D16 alloy.

For a more detailed study of the causes of crack formation in forgings special investigation was made of two forgings from the same melt after heat treating them as follows: heating to 495^{+50} , exposure for two hours, quenching in water heated to $60-80^{\circ}$. In one of these forgings, when ultrasonically inspected after physical treatment, cracks were discovered. In the other forging there were no cracks. The results of investigating the physical properties, chemical composition, microstructure, and fracture at the location of the crack showed that the content of alloying components and impurities in both forgings was practically the same.

The ultimate strength of the crack-containing forging is on the average $1-2 \text{ kg/mm}^2$ above the ultimate strength of the crackless forging. This difference can probably be explained by the fact that the forging with the crack was heated to a higher temperature than the crackless forging. On investigation of the forgings with and without cracks no non-metallic inclusions or other defects were discovered.

An examination of the forging fracture showed that the crack

formed in the sound metal.

When heated up to the quenching point in shaft furnaces without forced air circulation, the forgings are found to be in dissimilar conditions across a section of the charge. The peripheral forgings are more intensely bathed by the gases than those arranged in the middle part of the charge. Therefore the forgings on the edges may be heated to a higher temperature.

In order to prevent overheating of the peripheral forgings it is necessary strictly to maintain heating conditions in the furnace. This can be done by controlling the temperature directly at the metal.

Complete elimination of spoilage from cracks was achieved after introducing a method of step-quenching into batch production. A method of quenching of this sort applicable to AK₄ alloy is described in detail by V. I. Dobatkin*.

Step-quenching of D1 alloy forgings was conducted according to the following regime. The forgings, without physical pre-treatment, were heated for quench-hardening in a saltpeter bath to $500 \pm 5^{\circ}$ C and held for 30 minutes. The first cooling step took place in a fusion of salts at $145-155^{\circ}$ for 115 minutes; the second, in water at $50-70^{\circ}$. Eight forgings were loaded at the same time in the charge compartment.

In the forgings heated by the step method without physical pre-treatment the physical properties in two zones were investigated.

Specimens were cut out of the first zone longitudinally (specimen No. 1 in Fig. 1). In the second zone specimens were cut out of the lower compact part transversely (specimen No. 6 in Fig. 1). The

* V. I. Dobatkin, Step-Quenching Aluminum Alloys, in collection Legkyye Splavy, No. 1, AN SSSR, 1958.

frequency of recurrence of physical properties after ordinary and step-quenching of the batched forgings is shown in Fig. 2.

The physical test results show that the ultimate strength of the specimens cut from the forgings and processed by step-quenching is 2 kg/mm^2 less than forgings which underwent quenching in the ordinary way. The relative elongation of the specimens from the second zone amounts on the average to 6%, which corresponds to the degree of relative elongation of the forgings processed in the ordinary way.

Results of the research on the magnitudes of the residual stresses, using the physical method of N. N. Davidenkov, are shown in Table 3.

It must be mentioned that for a long time not one case of failure was observed in the physical processing of step-quenched forgings.

Thus a considerable decrease in the values of the residual stresses as a result of step-quenching made it completely feasible to eliminate forging-spoilage from cracks.

Conclusions

It has been ascertained that large residual thermal stresses appearing in quenching are the cause of spoilage of heavy D1 alloy forgings from cracks. These stresses are concentrated in areas of transition from fine profiles to heavy ones when circular grooves are lathe turned, causing cracking. Moreover, raising the heating temperature for quenching increases the tendency to cracking in the forgings.

To eliminate crack spoilage from forgings it is necessary to keep strictly to the temperature of heating for quenching and to use hot cooling media (water heated to 80° or a fusion of salts at $145-155^\circ$ in step quenching).

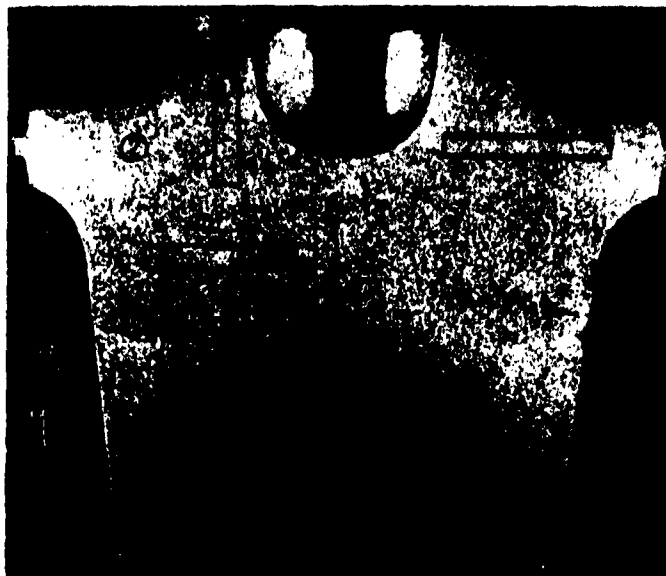


Fig. 1.

Diagram of specimens cut from forging
for physical tests
(specimens numbered 1-7)

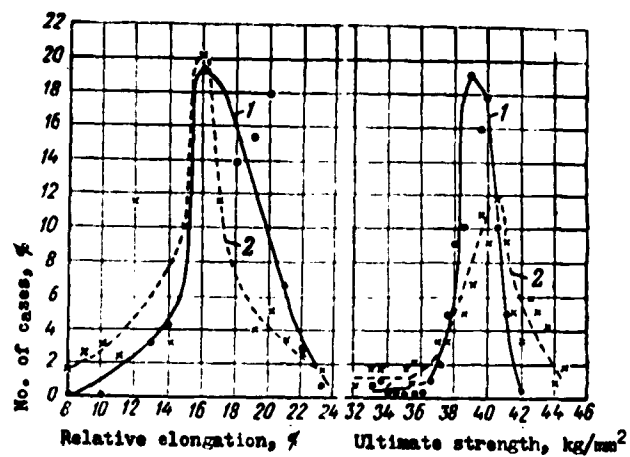


Fig. 2.

Frequency of recurrence of physical
properties in forgings (Longitudinal
specimens) after

1) Step quenching 2) Ordinary quenching

TABLE 1

Physical Properties of Forgings

| No. of Specimen (Fig. 1) | Direction of Cutting Specimen in Relation to Metal Flow | Ultimate Strength σ_b , kg/mm ² | Relative Elongation % |
|--------------------------|---|---|-----------------------|
| 1 | Longitudinal | 37,5 | 17,0 |
| 2 | Chord-wise | 41,8 | 18,0 |
| 3 | Longitudinal | 37,1 | 8,6 |
| 4 | Transverse | 36,5 | 12,0 |
| 5 | Longitudinal | 34,1 | 9,8 |
| 6 | Transverse | 31,8 | 4,6 |
| 7 | Longitudinal | 32,0 | 5,2 |

Notes: Specimen No. 1 is the control in batching forgings in accordance with STU [technical specifications] in batch production heat treatment of the forgings followed physical treatment. Forgings were heated for quenching in shaft furnaces without forced air circulation and with gas or oil as a fuel.

TABLE 2

Cracking in Forgings Depending on Type of Heat Treatment

| Temperature of Heating for Quenching, °C | Holding Period, Hours | Water Coolant temperature, °C | Crack Spoilage of Forgings, % |
|--|-----------------------|-------------------------------|-------------------------------|
| 490±5 | 2,0 | 20—30 | 3,9 |
| 495±5 | 4,0 | 30—40 | 13,5 |
| 495±5 | 4,0 | 60—70 | 1,5 |
| 490±5 | 2,0 | 60—80 | 0,2 |

TABLE 3

Maximum Residual Stresses in kg/mm²

| Method of quenching | Tangential | | Radial Tensile |
|------------------------|------------|-------------|----------------|
| | Tensile | Compressive | |
| In water heated to 50° | +1,0 | -19,7 | +6,7 |
| Step quenching | +3,4 | -9,2 | +1,8 |

HOT CRACKING OF ALLOYS OF THE V95 TYPE*

I. I. Novikov and A. E. Semenov

The tendency of an alloy to form shrinkage cracks depends above all on its chemical composition of alloying components and impurities. Controlling the silicon and iron content is widely used in industry for raising crack resistance in continuous ingot casting of aluminum alloys [1].

Controlling the content of other alloying components has been considerably less well studied and is altogether inadequately used to increase crack resistance. At the same time changing the composition of basic components even only within the limits of permissible variations of the All-Union [Soviet] State Standard may exert a substantial influence on the tendency of an alloy to form shrinkage cracks. Studying the influence of the chemical composition of alloying components and impurities on the tendency of industrial alloys to form shrinkage cracks, we ought to separate the influence of the composition on the formation of hot and cold cracks, since measures for combating these

* G. V. Indenbaum, A. V. Blinov, and I. S. Kirpa participated in the experiments.

two types of spoilage may be different.

Crystallization cracks forming in restricted contraction in solid solution it is expedient to call hot cracks [2-5]. For alloys with a considerable crystallization range the probability of the ingot's being affected by hot cracks is the greater, the larger the zone of hot cracking in the ingot [6]. The extent of the hot cracking zone in the ingot is determined by the size of the "effective" crystallization range, by the plasticity of the alloy in solid solution, by the conditions for remedying incipient cracks, by the rate of pour, by the height of the mold, and by other factors.

The chemical composition influences the hot cracking as a result of a change in the "effective" range of crystallization, in the plasticity of the alloy at temperatures higher than the solidus [7], and in the ability to "remedy" shrinkage cracks [8, 9].

In the present article the results of studying the influence of the chemical composition on hot cracking in alloys of the V95 type are scrutinized. The feasibility of obtaining large ingots of these alloys is limited chiefly by the formation of hot cracks.

Comparatively small changes in chemical composition (tenths of one percent) cannot as a rule substantially change the "effective" range of crystallization and the ability to remedy shrinkage cracks. Therefore the effect of small changes in chemical composition on hot cracking should be connected with the change in the plasticity of the alloy in solid solution.

In the investigation we conducted, the hot cracking was evaluated by the technological ring test. The ratio of the length of the basic crack to the perimeter of a radial section of the ring expressed in percent was the index of hot cracking. As the research of many authors

has demonstrated, this method characterizes hot cracking in an entirely satisfactory manner [10, 11].

Hot cracking of aluminum alloys in the ring test (outside diameter of the ring—60 mm, inside diameter—46 mm, height—20 mm):

| Alloy | V95 | D1 (0.2% Si) | D1 (0.5% Si) | AK6 | AK8 | A20 |
|--------------|-----|--------------|--------------|-----|-----|-----|
| Hot cracking | 100 | 95 | 20 | 42 | 3 | 2 |

In continuous ingot casting, alloys arrange themselves into a number of orders, shown in the table, according to their tendency to hot cracking.

In the research on the effect of chemical composition on hot cracking of type V95 alloys 15 to 20 rings of each composition were cast. The dimensions of the rings were the same for each test series, in which the content of one component or another was changed. These dimensions were chosen in order to show in a given test series the regularity of change in hot cracking on altering the content of one component. Since in the different test series rings of different sizes were used, it is impossible to compare with one another the absolute values of hot cracking for alloys from the different series. The casting temperature in all tests was 690°. The change in the content of an element in the alloy was produced by the preliminary preparation of the charge of the pertinent alloying element in the initial alloy of average composition (within the limits of the All-Union [Soviet] State Standard).

Tests for rupture in solid solution were accomplished according to the methods set forth in other works [12].

A study of the effect of manganese on hot cracking in V95 alloy (Fig. 1) shows that a small decrease in the manganese content of the

alloy sharply diminishes its tendency to hot cracking. Differential thermal analysis showed that a manganese content change in the alloy from 0.2 to 0.6% had practically no influence on the magnitude of the "effective" range of crystallization.

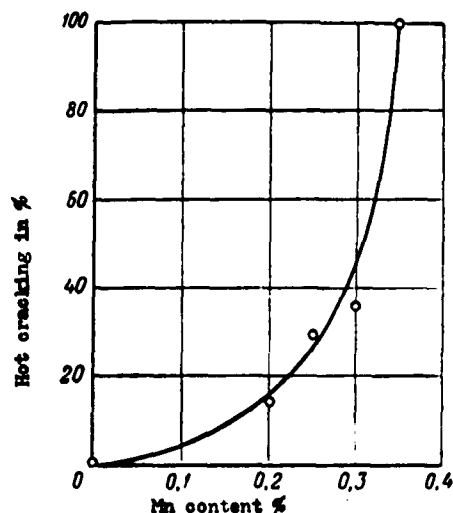


Fig. 1. Dependence of hot cracking in V95 alloy on manganese content.

The dependence of V95 alloy elongation per unit length on the temperature in the crystallization range at different manganese contents (Fig. 2) agrees with the results in other works [13, 14]. The curves in Fig. 2 are drawn by the method of least squares. It should be noted that the small difference in absolute value in elongation per unit length in the alloys compared

must have a considerable effect on hot cracking, since it is commensurate with the values of linear shrinkage in the "effective" crystallization range.

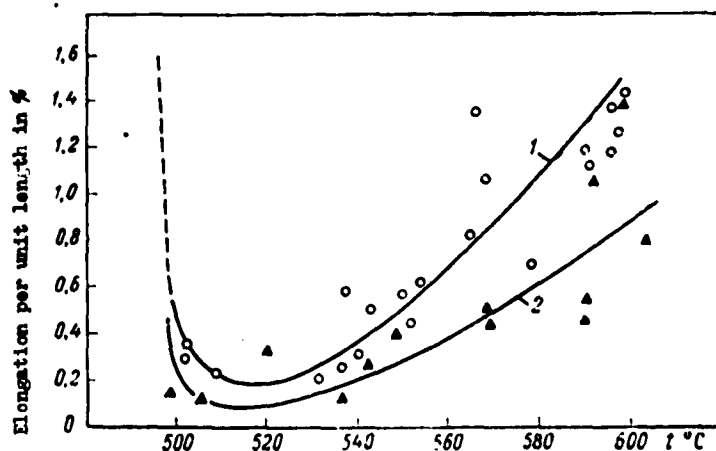


Fig. 2. Dependence of elongation per unit length in V95 alloy on temperature in solid solution with a manganese content of 1) 0.21% and 2) 0.44%.

These data confirm the rule established earlier by I. N. Fridlyander, V. I. Dobatkin, V. I. Khol'nova, and others that V93 alloy containing no Mn or Cr possesses a lesser inclination to hot cracking than V95 alloy which has Mn in its composition (and also Cr) [15].

Figure 3 shows the influence of silicon and iron on hot cracking. When silicon content is increased and iron content is decreased in V95 alloy, its hot cracking increases considerably. Analogous results were obtained by V. I. Dobatkin when casting ingots. The effect of silicon and iron on crack resistance is caused by the action of these alloying elements on the alloy's plasticity in solid solution (Fig. 4). With a high iron and a low silicon content the region of little plasticity spreads to the smallest temperature range and in connection with this the zone of hot cracking in the ingot diminishes.

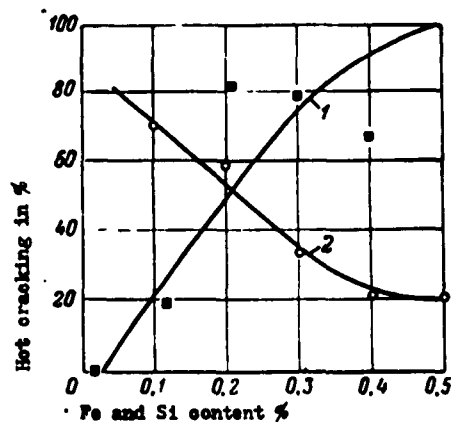


Fig. 3. Dependence of hot cracking in V95 alloy on content of 1) silicon and 2) iron

Of the other components magnesium exerts the most considerable influence on hot cracking in V95 alloy (Fig. 5).

Changing the zinc and copper content within the tolerances of the All-Union [Soviet] State Standard had no influence on hot cracking in V95 alloy.

Conclusions

Changing the manganese, iron, silicon, and magnesium content in alloys of the V95 type may substantially decrease their tendency to form hot cracks in continuous ingot casting.

The effect of manganese, iron, and silicon on hot cracking in type V95 alloys is connected with a change in their plasticity in solid solution in continuously cast ingots.

By the technological ring test one may in rough approximation evaluate the effect of the changes in chemical composition on the tendency to hot cracking in industrial aluminum alloys during continuous casting.

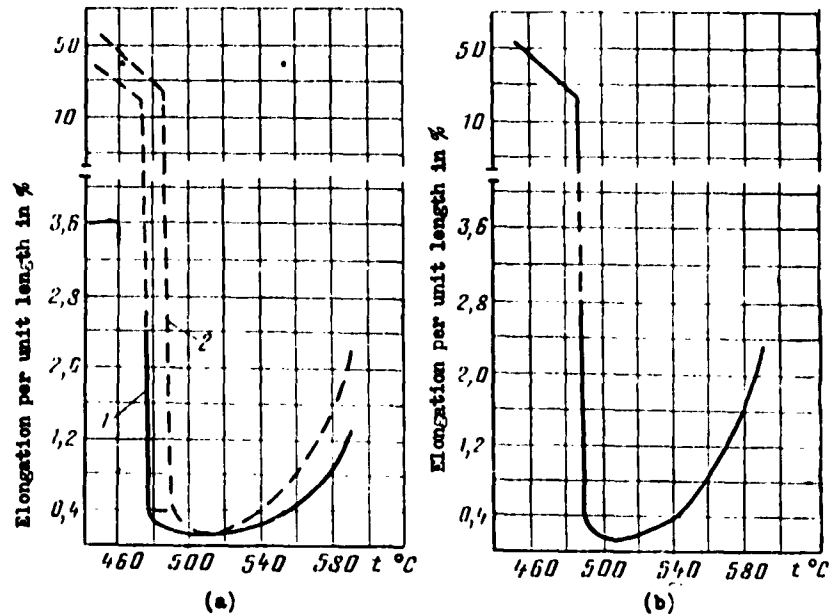


Fig. 4. Dependence of elongation per unit length in V95 alloy on temperature in solid solution with 0.1% silicon content and iron content of 1) 0.1% and 2) 0.55%

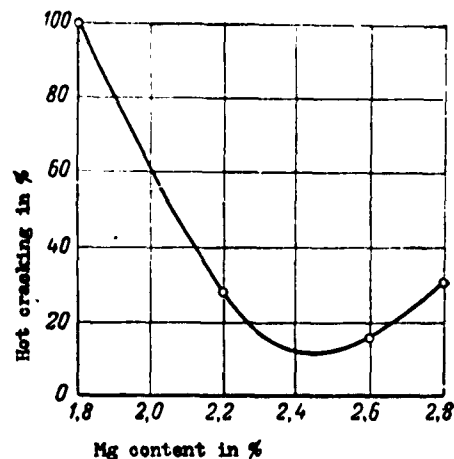


Fig. 5. Dependence of hot cracking in V95 alloy on magnesium content

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SOME FEATURES OF OBTAINING INGOTS FROM THE
ALLOYS AMg5V AND AMg6

V. A. Livanov, R. M. Gabidullin
and L. P. Donorskaya

The technology of casting ingots of aluminum alloys containing more than 5% of magnesium has certain peculiarities.

For the production of sheets and plates of AMg5V and AMg6 alloys flat ingots are continuously cast through a mold 205 x 1400 mm having notches on the narrow faces. The pouring temperature is within the limits of 670-680° and the drawing rate of the ingots amounts to 65-75 mm/min. For protection from bottom cracks the aluminum is poured on a bottom-plate stool.

Experience has shown that in such a casting regime ingot failure comes chiefly as a result of the formation of hot surface cracks. The spoilage in the individual melts from this amounted to 70%. The tendency of high magnesium alloys to form hot cracks is explicable by the slight strength of the alloy at high temperature and by the oxide coating, which forms on the surface of the ingot, being composed of magnesium oxide [1]. Oxidation of the magnesium is accompanied by a decrease in volume (the amount of magnesium oxide is 58% of the

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metallic magnesium [2]; and the oxide coating on the surface of the ingot is consequently friable and loose. Microcracks appearing on the surface become strain concentrations in spots and, when not uniformly enough cooled, cause the appearance of hot cracks.

Therefore the work was done with the aim of preventing hot cracks and also of diminishing the general hot cracking of alloys and of changing the conditions found in the formation of the oxide coating (increasing its density and strength).

The effect of the different components in the alloy on its hot cracking was ascertained by processing the statistical data on casting ingots of AMg5V and AMg6. Analysis of the results obtained showed that a change in the magnesium content within the limits of 4.8-7.0% has practically no effect on hot cracking; increasing the manganese content from 0.5 to 0.75% increases spoilage from superficial hot cracks up to 15%; increasing the iron content from 0.15 to 0.25% lowers the quantity of spoilage; increasing the ratio of the iron to silicon content also considerably lowers the amount of ingot spoilage from superficial hot cracks.

The test melts then made confirmed these patterns.

The effect of the iron additive may be explained on examining the quasi-binary section Al-Mg₂Si. With an Mg₂Si content greater than 1.85%, i.e., with a silicon content of more than 0.67%, the eutectic appears, but in quantities still insufficient to fill the intercrystalline shrinkage gaps. With a larger magnesium content than is necessary for the formation of the compound Mg₂Si the solubility of the Mg₂Si in the solid solution falls to zero [3]. In connection with this the eutectic may appear even when there is an insignificant silicon content. This amount is insufficient, however,

to fill the shrinkage gaps; and hot cracking of the alloy sharply rises. Introducing iron into the alloy lowers hot cracking and this is evidently explained by the appearance of the chemical compound $Al_xFe_ySi_z$.

The influence of the manganese is probably connected with a decrease in the alloy's ductility at low temperatures and temperatures of crystallization.

In the work of M. V. Mal'tsev and Yu. D. Chistyakov mentioned above it is pointed out that oxidation of the components in aluminum alloys has a selective nature. When 0.005-0.05% Be is introduced into an aluminum-magnesium alloy the oxide coating has a heterogeneous nature and consists of beryllium oxide and magnesium oxide. This coat possesses high density and strength; and that the beryllium atoms, penetrating into the pores and cracks of the magnesium oxide coating, fill them and "cure" them by themselves oxidizing, explains this.

The pouring of test melts of MAG5B and AMg6 alloy with an additive of beryllium in quantities of 0.003-0.005% showed the good quality of the ingot surface, but the high cost of beryllium limits its use in mass production.

Investigations made subsequently showed that to obtain a high-grade ingot surface an additive of 0.0001-0.0002% Be was sufficient; with this the hot cracking of the alloys also sharply decreased.

The beryllium additive changes the outward appearance of the ingot unrecognizably. The ingot acquires a silvery color instead of the dark one when there is no beryllium in the alloy.

When the ingots are poured, intermetallic inclusions of complex composition are a spoilage specific for AMg5V and AMg6. Analysis has shown that the cause of their appearance in the structure is the

presence in the alloy of little-soluble additives of titanium and vanadium. A treatment of the statistical data on the chemical analysis of a melt prepared with different calculated compositions of titanium and vanadium showed that the quantity of titanium and vanadium contained in the alloy did not depend, in the limits investigated, on the calculated composition. To establish the dependence of the titanium and vanadium content in the alloy on the temperature of fusion, a series of melts of AMg5V and AMg6 alloys was made in a laboratory furnace of 5kg capacity according to the calculated composition in the table:

| Make of Alloy | Calculated Composition in % | | | | | |
|---------------|-----------------------------|------|-----|-----|-----|-----|
| | Mg | Mn | Ti | V | Fe | Si |
| AMg6 | 6.5 | 0.65 | 0.3 | — | 0.2 | 0.2 |
| AMg5V | 5.2 | 0.45 | — | 0.3 | 0.2 | 0.2 |

The metal was overheated to 800°, carefully stirred, and kept at this temperature for 1 hour, after which samples for chemical analysis were taken from the upper part of the liquid metal. Then the temperature was lowered, the melt was again kept there for an hour, and a sample was taken anew for chemical analysis. Similarly, samples were taken out down to 650°.

The results of chemical analysis of the extracted samples (see figure) show that with the decrease in temperature the solubility of titanium and vanadium in the liquid metal declines. Thus at a high calculated composition of titanium and vanadium in the alloy they will not assimilate and losses of these components in the slag will be very high. On feeding the metal from the mixer into the mold the

the temperature falls; and, in case there is a high content of titanium and vanadium, crystallization of intermetallic compounds begins in the feeding equipment. V. I. Dobatkin, whose recommendations should be observed when pouring ingots of AMg5V and AMg6 alloy, studied in detail [4] the conditions of intermetallic compound growth in a hole in the liquid metal. It should be pointed out that intermetallic compounds appear considerably more often in pouring ingots of large diameter.

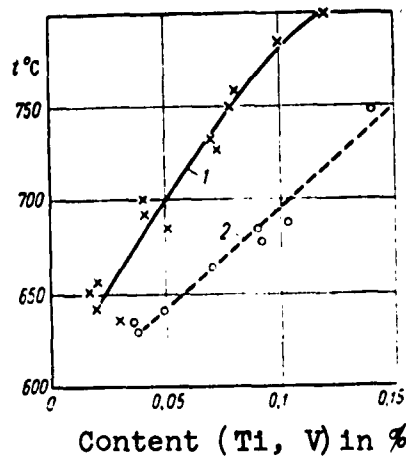


Fig. 1. Change in solubility depending on the temperature: 1) vanadium in AMg5V alloy, 2) titanium in AMg6 alloy.

As V. I. Dobatkin indicated, the growth of intermetallic compound crystals in a hole occurs in controlled overcooling Δt . Of interest is the mathematical analysis of the appearance of intermetallic compounds as a function of the ingot diameter. Taking the rate of crystal growth $C=C(\Delta t)$, we note that when the temperature in the hole is constant, the rate of crystal growth C must be constant, since $\Delta t = \text{const}$. The size of the crystals after time τ will amount to

$$\delta = C\tau, \quad (1)$$

The time during which crystal growth is possible may be defined as

$$\tau = \frac{h}{V_p}, \quad (2)$$

where h is the height from the surface to the crystallizing layer;
 V_p , the rate of pouring;
 but

$$h = \frac{R-r}{R} h_h \quad (3)$$

where R is the radius of the ingot and h_h , the depth of the hole.

The depth of the hole may be determined by V. A. Livanov's formula [5]

$$h_h = V_p(\gamma + \beta H), \quad (4)$$

where γ and β are factors depending on the diameter of the ingot and the construction of the mold; and H is the height of the level of the ingot in the mold.

From (1), (2), (3), and (4) we will get

$$\delta = k\left(1 - \frac{r}{R}\right), \quad (5)$$

where

$$k = C(\gamma + \beta H). \quad (6)$$

When the ingot diameter is changed the factor γ changes in accordance with the following law:

$$\gamma = \gamma_0 R^2, \quad (7)$$

where γ_0 is a constant factor.

From (5) and (7) we get

$$\gamma = C\left(1 - \frac{r}{R}\right)(\gamma_0 R^2 + \beta H). \quad (8)$$

Analyzing the last formula we may note the following:

the dimensions of the crystallites of intermetallic compounds do not depend at a given diameter on the rate of pouring;

the dimensions of the intermetallic compounds change from the edge of the ingot to its center in accordance with a rectilinear law;

with a change in the diameter of the ingot the dimensions of the intermetallic compounds (and, consequently, also the quantity observable visually) change proportionally to the square of the ingot diameter, and this is confirmed by the experimental data;

the dimensions and quantity of the observed intermetallic compounds change proportionally to the height of the level of the liquid metal in the mold.

Conclusions

To decrease hot cracking in AMg5V and AMg6 alloy it is necessary to:

- keep the iron content-silicon content ratio higher than 1.5;
- keep the manganese content of the alloy at 0.50-0.55%;
- prepare the beryllium charge in the amount of 0.0001-0.0002%.

To obtain ingots without gross accumulations of intermetallic compounds the titanium and vanadium content in AMg6 and AMg5V alloy must be 0.02-0.05%.

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